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A MODEL FOR TENSILE FRACTURE OF CARBON-CARBON COMPOSITE FIBER BUNDLES

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ABSTRACT

A probabilistic model for tensile fracture of straight-fiber carbon-carbon composites is proposed. The analysis derives from the extensive theoretical work available for graphite/epoxy composites (eg, Batdorf, Manders et al, Rosen, and Phoenix) but attempts to account for the weak microcracked interfaces in carbon-carbons by assuming load transfer between fiber and matrix is primarily frictional. The model extends the work of Chatterjee, et al, by including a predictive analysis for the frictional shear stress, incorporating a Poisson's effect (from Gent) and a thermal-expansion effect. Inputs to the model include fiber, matrix, and interface properties (including friction coefficient), fiber strength distribution or the length dependence of dry-yarn strength, and transverse stresses acting on the yarn bundle. Illustrative results show that composite strength may be expected to increase with temperature even if the fiber strength does not. Also, the results show that room-temperature strength of a carbon-carbon yarn tends to be significantly lower than the strength of a similar graphite-epoxy yarn, even if no degradation of fiber properties has occurred during the fabrication of the carbon-carbon composite.

INTRODUCTION

The effective strength of fibers in a carbon-carbon can be less than the average strength of the fibers. For a straight-fiber composite (eg, unidirectional, 2D tape laminate, or a Cartesian 3D block), the fiber utilization factor (or fiber strength efficiency) may be expressed as:

$$\eta = \frac{F_{ct}}{v_f \bar{\sigma}_f} \tag{1}$$

where F_{ct} is the composite tensile strength in the fiber direction, v_f is the fiber volume fraction in that direction, and $\bar{\sigma}_f$ is the average fiber strength. Omission of matrix strength from Eq 1 appears reasonable for typical carbon-carbons (v_f = 10 to 65 percent, approximately) because the matrix strength is much less than the fiber strength. Taking the values of $\bar{\sigma}_f$ as they are reported by the fiber manufacturer, Table 1 provides some examples of fiber utilization factors for several carbon-carbon composites. These strength efficiencies are generally low, some being less than fifty percent. We might be tempted to ascribe the low efficiencies to "degradation" of the fibers during fabrication of the composite, or to misalignments of the fiber axis from the load axis (as in the undulations of yarns in a woven cloth composite). However, in judging whether or not fibers have been degraded or whether misalignment is a significant factor, it is necessary to estimate the strength of a similar straight-fiber composite in which no degradation has taken place.

Using a probabilistic approach, Chatterjee et al (Ref. 1) analyzed the on-axis tensile strength of fine-weave 3D carbon-carbons, made of T-50 (rayon) fibers, and concluded that the observed strength was approximately equal to the expected strength of a composite made with fibers having the strength variability observed in T-50 fibers and the weak interfaces typical of carbon-carbons. That is, the strength efficiency factor of about 60 percent (Table 1) for these composites can be explained in terms of the statistics of undegraded fibers. A similar conclusion would be reached for HM fibers in fine-weave 3D composites. However, the very low strength efficiency of carbon-carbons made with T-300 yarns (Table 1) probably cannot be attributed entirely to probabilistic effects. In such cases, the application of probabilistic theory would provide a baseline from which to estimate the extent of fiber degradation induced during fabrication of the composite.

We should note that the analysis of Chatterjee et al is incomplete in the sense that they backed into their conclusion by assuming no fiber degradation and finding that the consequences, in terms of certain initially unknown parameters (chiefly the strength of the fiber-matrix interface) used in their probabilistic model, appeared reasonable for carbon-carbons. In the probabilistic approach proposed here, an attempt is made to make analytical estimates of the fiber-matrix interfacial shear stress.

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ANALYTICAL BACKGROUND

At tensile fracture of a uni-directional composite the average stress carried by the fibers can be different than the average strength measured on fibers alone. In the absence of a difference between in-situ and dry fiber properties, the average fiber stress at composite fracture differs from the average dry-fiber strength because of factors best described in probabilistic terms. Basically, there are two effects: a "series" effect expressed in terms of weakest-link theory; and a "parallel" effect that deals with load re-distribution among the fibers upon fracture of one or more fibers in a bundle. These effects are well described in Ref. 2; a synopsis is given below.

The series or weakest-link effect is illustrated by the experimental fact that fiber strength measured in single-fiber tests using a long gage length is lower than in tests on a short gage length. This can be understood by considering a long length to be made of a chain of short lengths, and invoking the chain rule of probability:

$$p_{sL} = \left(p_s\right)^L \tag{1}$$

where p_s is the probability of survival of a specimen of unit length, and p_{sL} is the probability of survival of a specimen of length L, both at the same applied stress. Thus, if the effective length of fibers in a composite specimen is different than the gage length used to generate the fiber strength distribution, the average strength of fibers in the composite may differ from the average measured fiber strength.

The problem of estimating the effective length of fibers within a composite relates to load redistribution. If the matrix is ineffective (or absent) and therefore cannot transfer load from one fiber to the next over a finite length, the fracture of one fiber in a bundle of N fibers means that each of the remaining unfractured fibers carries an increased load:

$$\sigma' = \bar{\sigma}\left(1 - \frac{n}{N}\right)^{-1} \tag{2}$$

where σ' is the new fiber stress, n is the number of broken fibers, and $\bar{\sigma}$ is the fiber stress that would be calculated if all the fibers were carrying the load. This situation corresponds to the strength of dry or "uncoupled" fiber bundles, which has been treated theoretically by Coleman (1958, cited in Ref. 2) for fibers having a strength distribution that can be described by a two-parameter Weibull equation:

$$p_{sL} = \exp\left(-L\left(\frac{\bar{\sigma}}{\sigma_0}\right)^m\right) \tag{3}$$

where m is the Weibull shape factor and σ_0 is the scale factor, and p_{sL} is the survival probability at stress $\bar{\sigma}$. For this probability distribution, the mean strength of fibers, σ^*, is:

$$\frac{\sigma^*}{\sigma_0} = L^{-\frac{1}{m}}\,\Gamma\left(1+\frac{1}{m}\right) \tag{4}$$

where the gamma function ranges from a minimum of about 0.88 (for m = 2) to about 0.98 for m = 30. The strength of uncoupled bundles is related to the average strength of the fibers (of the same length) by the Coleman factor:

$$\epsilon = \left[m^{1/m}\exp\left(\frac{1}{m}\right)\Gamma\left(1+\frac{1}{m}\right)\right]^{-1} \tag{5}$$

In a composite having an effective matrix, such that load is transferred from a broken fiber (or group of broken fibers) preferentially to the immediately neighboring fibers, the effective length of the fibers may be considerably shorter than the specimen gage length. Rosen (Ref. 3) estimates this effective length in terms of the shear lag distance in the matrix, and treats the composite as a chain of minibundles each having this effective length. Within each minibundle, the load redistribution is assumed to be uniform; that is, Rosen assumes each minibundle is an uncoupled bundle of length δ. The strength of the composite is then taken to be the strength of an uncoupled bundle of length δ; because δ is in general shorter than the composite's gage length L_g, the strength of the composite is higher than the strength of dry bundle of length L_g. Recent treatments of the problem by Batdorf, Phoenix et al, and Manders et al (Ref. 4, 5, and 2), among others, consider the effects of increased stress within fibers that border the broken fiber(s). They abandon the concept of a chain of dry mini-bundles, and adopt instead the concept that composite fracture occurs when fibers in the vicinity of a multiple fiber break are likely to fracture under no increase in applied stress. This approach, which may be termed as dealing with "coupled" bundles, also relies on the effective-length concept by assuming that the fibers next to a break experience higher stresses only in a region of finite length defined by the shear lag

TABLE 1. STRENGTH EFFICIENCIES OF SOME CARBON-CARBON COMPOSITES @ ROOM TEMPERATURE

FIBER	AVERAGE FIBER STRENGTH KSI (approx.)	COMPOSITE TYPE	AVERAGE FIBER STRENGTH IN-SITU KSI (approx.)	STRENGTH EFFICIENCY FACTOR	SOURCE FOR COMPOSITE DATA (see notes)	REMARKS
T-50 (rayon)	315	3D block	190	0.60	A	fine-weave Cartesian
HM (PAN)	340	3D block	215	0.63	B	fine-weave Cartesian
T-300	400	3D cylinder	100	0.25	C	various woven cylinders
T-300 heat-treated	350	3D cylinder	100	0.29	C	same woven billets, treated T-300 assumed equivalent to T-50 PAN
T-300 heat-treated	350	2D laminate (8-harness satin cloth laminate)	160	0.46	D	ACC-4 w/o inhibitors, warp direction
WYB (rayon)	90	2D laminate (plain-weave cloth laminate)	21	0.23	E	KKARB 1200 involutes, warp direction
Japanese (PAN?)	286	unidirectional 50 % vol. fraction	110	0.38	F	experimental composite with furfural alcohol based matrix
Japanese (PAN?) treated to 2800 C	146	unidirectional 50 % vol. fraction	110	0.75	F	same composite as immed. above referred

Fiber strength data from compilations from Ref. G, Ref. H and Union Carbide literature.
Japanese PAN data from Ref. F.

In-situ fiber strength estimated from on-axis composite strength using estimated fiber volume fraction.

Sources for data are:

A. Littleton, H. E., and Pears, C. D., Mechanical, Thermal and Nondestructive Characterization of GE-2.2.3, AFML-TR-77-48, April 1977.

B. Pears, C. D., et al, Evaluation of C-C Composites for the P4 Program, AFML-TR-78-2, Feb 1979.

C. Kibler, J. J., In-Situ T-300 Fiber Properties in Tension and Compression in Carbon-Carbon Cylindrical Weaves, 1983 JANNAF RNTS Mtg, Colorado Springs, Dec. 1983.

D. Starrett, Stuart, Prelim. Data for Structural Carbon-Carbon Composites, SoRI-EAS-8[illegible], Dec 1983.

E. Davis, H. O., and Vronay, D. F., Structural Assessment of Involutes, AFML-TR-79-4068, June 1979.

F. Yamada, Shigehiko, and Tamada, Koshi, Graphitization of Carbon-Fibre/Glassy Carbon Composites, J. Comp. Mat., reprinted in Carbon Composite and Metal Composite Systems, C. J. Hilado, Ed, Technomic Publishing Co, 1974, pp18-2[illegible].

G. Schmidt, D. L., Replacement Fibers for Thermal Protection Applications, AFML-TR-77-68, Aug 1977.

H. Chard, W. and McCall, J., Assessment of the Availability and Utilization of Carbon-Base Fibers for DoD Applications, Part I - Final Report Summary, Batteile Columbus Labs, December, 1977.

characteristics of the fiber-matrix interface. Because of the stress concentrations surrounding a fiber break, the composite strength is lower than would be predicted by the Rosen model for the same effective length. For the composites considered by Manders et al, the difference between the strengths predicted by the coupled-bundle model and the Rosen model is small (about five percent, Ref. 2).

Applying the coupled bundle analysis to new composites requires estimates of the effective length and the stress-concentrations. For graphite/epoxy composites, in which the matrix and interface bond are reasonably effective, the effective length has been estimated to be in the vicinity of 10 fiber diameters, and the peak stress in near-neighbor fibers is estimated to be about 25 percent higher than the average fiber stress (Ref. 2). The effective length increases if the matrix debonds locally from the broken fiber; the debond length is undoubtedly a function of applied stress, and the stress concentration is probably lower for longer debonds. Given the weakness of graphite matrices in shear, and the ubiquitous microcracking observed in composites that have been processed at high temperatures, it is likely that the effective lengths for carbon-carbon composites are very much longer than for epoxy-matrix composites. Chatterjee et al (Ref. 1) have estimated the effective length in 3D carbon-carbons to be of the order of 1 inch, which translates to more than a thousand fiber diameters; they treat the load transfer in a carbon-carbon as being frictional (across a debonded interface) and the effective length δ as being proportional to the inverse of the frictional shear stress τ :

$$\delta = \frac{\bar{\sigma} r_f}{\tau} \cong \frac{\sigma_c r_f}{\tau V_f} \qquad (6)$$

where $\bar{\sigma}$ is the average fiber stress, σ_c is the stress applied to the 1D composite, V_f is the fiber volume fraction, and r_f the fiber radius. The shear lag distance over which elastic stress transfer occurs can be shown small in relation to Eq 6, if τ is small relative to the interface shear strength and the fiber stress is high (Ref. 1).

Given the small difference between coupled bundle predictions of composite strength and the predictions of the Rosen model (Manders et al, Ref. 2), and the decrease in coupling (stress concentration) that occurs as effective length increases, it is probably adequate to avoid the complications of coupled-bundle theory and use the uncoupled Rosen approach in treating carbon-carbons. Thus, being ignorant actual stress concentration factor is relatively unimportant, and the chief unknown becomes the frictional shear stress (Eq 6). This is the basis for Chatterjee et al's analysis (Ref. 1), shown schematically in Figure 1.

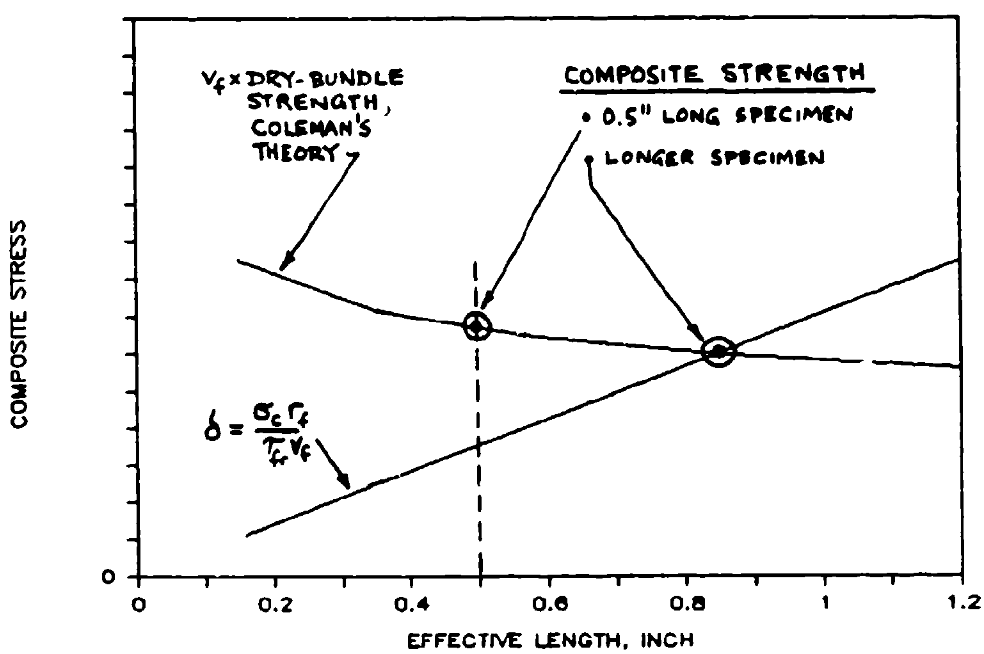


Figure 1. Schematic of the tensile fracture model of Chatterjee et al (Ref. 1).

The graph of Figure 1 is based on the use of Equation 6 to predict effective lengths that are linear with applied stress, assuming a constant value for the frictional shear stress. Fracture of the composite is predicted at an applied stress large enough that the effective length becomes such that the uncoupled-bundle strength is equal to the average fiber stress. An approximate limit to the effective length is the gage length L_G of the composite tensile specimen; if this limit is reached first, the composite strength estimate is the dry-bundle strength for bundles of length L_G.

If the fiber strength distribution is known and complies well with the Weibull equation (Eq. 3), the composite bundle strength curve in Figure 1 can readily be constructed from the Coleman analysis (Eq 5), the weakest link scaling law (Eq 1), using the volume fraction of fibers as the factor relating dry bundle strength to composite bundle strength. Alternatively, the composite bundle strength curve can be obtained from directly measured strengths of dry bundles of various lengths (as was done in Ref. 1); again, the composite bundle strength is the dry bundle strength multiplied by the fiber volume fraction.

NEW ANALYSIS

We proceed to extend the approach illustrated in Figure 1 by estimating the frictional shear stress as a function of the properties of the fiber, matrix, and their interface.

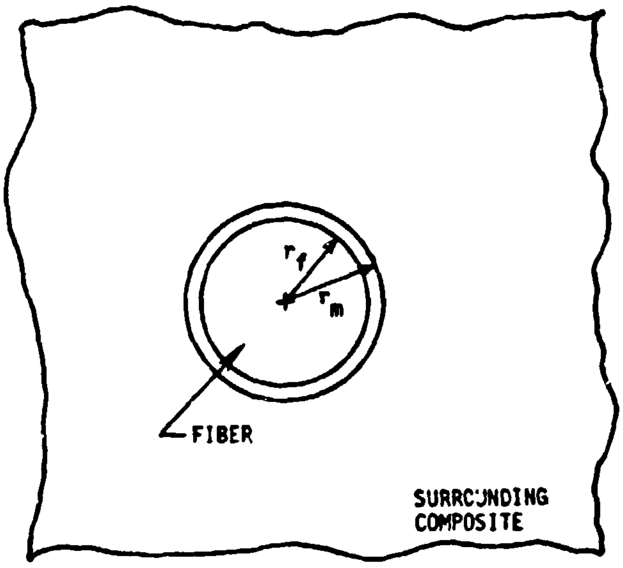
Assuming simple friction, the shear stress would be the product of the compressive radial stress across the interface and the friction coefficient. The radial stress could arise from two effects: a temperature-dependent compressive stress arising from the thermal-expansion mismatch between the fiber and the surrounding composite, plus a compressive stress that results from the Poisson's expansion of the fiber radius when the axial fiber stress is released; the concept of the Poisson's effect is suggested by the work of Gent (Ref. 6, for example).

The analysis proceeds by considering a (virtual) gap at the fiber-matrix interface (schematically shown in Fig. 2). On loading the composite in tension to stress σ_c the outer radius of the gap shrinks by Poisson's effect:

$$\Delta r_m = -\nu r_m \frac{\sigma_c}{E_{cL}} \tag{7}$$

where E_{cL} is the axial Young's modulus, and ν the axial Poisson's ratio, of the uni-directional composite . Under loading, the fiber radius also shrinks. However, when the fiber breaks its radius springs back to the original unloaded value, r_f . The change in gap, at the broken end of the fiber, is then:

$$\frac{\Delta g_e}{r_f} \simeq \frac{\Delta r_m}{r_m} = -\nu \frac{\sigma_c}{E_{cL}} \tag{8}$$



$g = r_m - r_f$

$\frac{g}{r_f} \ll 1$

Figure 2. Idealization of fiber-matrix interface.

assuming that the gap g is small in relation to the fiber radius. The gap is also affected by heating the composite, as a result of the difference between transverse expansivity of the fiber α_{fT} and that of the composite α_{cT}:

$$\frac{\Delta g_T}{r_f} \doteq \left(\alpha_{cT} - \alpha_{fT}\right) \Delta T \qquad (9)$$

where ΔT is the temperature rise. The net change in gap, at the broken end of the fiber, is:

$$\Delta g = \Delta g_P + \Delta g_T \qquad (10)$$

To account simply for the gradient in fiber tensile stress as frictional shear introduces load near the broken end, we use an estimate of the average gap change $\overline{\Delta g}$:

$$\overline{\Delta g} \doteq \frac{1}{2} \Delta g_P + \Delta g_T \qquad (11)$$

This simple averaging assumes that the Poisson's ratio of the fiber is approximately equal to that of the composite bundle. The effect of this average gap change on radial stress across the fiber-matrix interface may be estimated (by analogy to a cylindrical shrink fit, Ref. 7, for example) to be, approximately:

$$\overline{\sigma}_R \doteq \frac{\overline{E}_T \overline{\Delta g}}{2 r_f} \doteq \frac{\overline{E}_T}{2} \left[\left(\alpha_{cT} - \alpha_{fT}\right) \Delta T - \frac{\nu \sigma_c}{2 E_{cL}} \right] \qquad (12)$$

where $\overline{E}_T$ is an appropriate average of the composite's and the fiber's Young's moduli transverse to the fibers.

The total effective radial stress we use is the sum of $\overline{\sigma}_R$ from Eq 12 and an initial radial stress σ_R'. The initial radial stress may be estimated as the sum of any residual transverse stress (arising from cooldown from the last process heat-treatment) plus any external tractions imposed on the yarn bundle (from stresses applied to the composite or from mini-mechanical interactions within a multi-directional composite). Thus, the frictional shear stress capability is:

$$\tau = -\mu \left(\overline{\sigma}_R + \sigma_R' \right) + \tau' \qquad (13)$$

where μ is the coefficient of friction and τ' is a minimum frictional resistance due to other factors such as mechanical interlocking of micro-roughened interface surfaces.

The friction coefficient might be estimated from the extensive data available for graphites under friction (eg, Ref. 8 and 9). Unfortunately, the friction coefficient is affected significantly by various factors that are difficult to quantify for the composite, including adsorbed gases and the crystallographic nature of the surfaces. Thus, in the absence of direct data, we must guess the friction coefficients to implement this analysis.

The new analysis is the same as that of Chatterjee et al (Ref. 1) except that the shear stress used in Eq 6 is given by Eq 13, rather than taken as an arbitrary constant.

DISCUSSION OF LIMITATIONS

As noted in the development, the analysis is quite simplified. This is appropriate now because the experimental data (eg, Poisson's ratios of the fiber) necessary to support a refined analysis is not available. Also, by restricting the analysis to frictional shear lag, we have ignored the elastic shear lag. For the long effective lengths expected at low temperatures, this may be an appropriate approach. However, for some cases, it may be important to extend the model to include the elastic shear lag also.

If the fiber strength distribution cannot be well fitted by the Weibull distribution, two approaches may be considered. Neither seems difficult to implement. The first approach would be to determine whether the low-probability-of-fracture "tail" of the strength distribution may be well fitted by a Weibull curve. If so, the analysis would proceed using the local distribution, with some obvious modifications to the Coleman equations. The rationale for doing so derives from the fact that a minor fraction of fibers in a bundle need to break before the bundle breaks (Ref. 2); thus, only the low-probability tail of a distribution and the average fiber strength need be known with any accuracy to predict bundle failure. If the tail of the distribution does not conform well to a Weibull curve, it would be desirable to use the fiber strength distribution, directly as measured, by applying the chain-rule to shift it to the effective length, and by repeating Coleman's analysis numerically using the experimental distribution to predict bundle failure. Or, as pointed out earlier, data from strength tests on dry bundles of various lengths may be used directly.

The modeling described above assumes that the fiber bundle and a weak quasi-isotropic matrix are the constituents of the uni-directional carbon-carbon composite. This simplification ignores the possibility that a highly oriented matrix "sheath", which has been observed by microscope in several carbon-carbons (eg, Ref. 10 and 11), contributes to the composite strength. The role of the sheath is not well understood at present. Evangelides, Ref. 11 and 12, attributes to the sheath the observed fact that the effective Young's modulus of T-50 fibers in-situ is significantly greater than the virgin fiber modulus. Similar increases in effective fiber stiffness occur in composites made with other graphite fibers (eg, Ref. 13). However, factors other than the sheath may also contribute to the in-situ stiffening of graphite fibers; these include the fact that in-situ fiber density is higher than virgin fiber density, that the fibers undergo some stretching during the fabrication of carbon-carbons (Ref. 14 and 15), and that fiber properties are affected by temperature cycling (eg, Ref. 16); reasonable estimates of these effects can account for the magnitude of the stiffening (Ref. 14) without invoking the sheath effect.

Thus, it is not clear that the sheath actually plays the important role ascribed to it by Evangelides. The fact that the in-situ stiffness of pitch-densified 3D composites made with T-50 and T-75 fibers is the same (Ref. 13), in spite of a large difference in virgin-fiber stiffnesses, suggests that the sheath does not contribute significantly to the composite stiffness. At first glance, this conclusion may seem contradicted by the observed sheath-like orientations of matrices within c-c fiber bundles. However, when we consider the concurrent observation that the in-situ matrix is extensively microcracked, it does seem reasonable to discount its contribution to composite stiffness and strength, as in the simplified strength model described above.

As the uni-directional strength model depends heavily on good estimates of the interfacial friction, review of any available and relevant data is needed; also, new tests are recommended. Available data for quantifying the interfacial shear include experiments conducted in "microshear punch tests" on uni-directional composites (Seibold et al, Ref. 17) and on 3D and 4D composites (Loomis et al, Ref. 18). The test and representative load-deflection data are schematized in Fig. 3. In Seibold et al's tests, the punched-out plug is simply a portion of the uni-directional composite pushed out in the axial direction by the punch. In Loomis et al's tests, the punched-out plug is a complete yarn bundle pushed out of the surrounding 3D or 4D composite. While neither test directly deals with the fiber-matrix interface, the data is indicative of the general magnitude of the interface strengths in such composites. In the absence of a fiber pull-out or punch-out test, which would be difficult to perform because of the small dimensions of fibers, this seems as close as we can reasonably approach the required properties by experimentation.

The load-deflection traces in both Ref. 17 and Ref. 18 show a post-ultimate load that may be attributed to friction. Unfortunately, the friction observed in these tests includes friction between the punch and the composite, and the friction attributable to the plug-composite interface cannot be unambiguously derived from these tests. Indeed, the implied values of frictional shear stress are several hundred psi at room temperature, which (see Table 2) would imply rather (unreasonably?) short effective lengths in the composite.

To provide better friction data, a modified microshear test procedure has been suggested (Ref. 19). To avoid the punch-composite friction, the tests should involve changing to a slightly smaller diameter punch just after peak load is reached. In this way, the friction force will be due solely

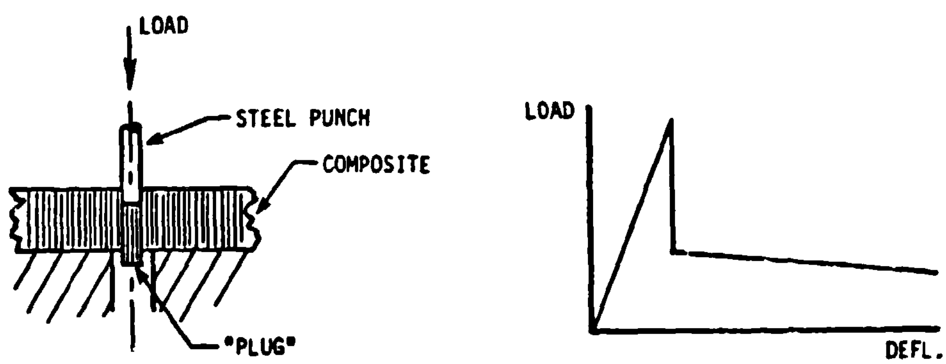


Figure 3. Schematic of the microshear punch test.

to plug-composite effects. An alternate approach would be to evaluate data from yarn pullout tests (such as are planned at UCLA, Ref. 22, or such as those described in Ref. 23). The analysis of punch-out data should account for the transverse compressive stress, generated by Poisson's expansion of the plug resulting from the axial punch force. An inverse Poisson's effect would occur in pull-out tests. The Poisson's effect may be treated in essentially the same manner as described above for the fiber-matrix friction, with consideration for the large difference between the Poisson's ratios of 3D composites and their yarns.

Experimental verification of the analytical model for tensile strength would include obtaining data regarding the effective length of the fibers within a composite. In the absence of direct data, effective length information most commonly is indirectly derived by comparing the behavior of the composite to the behavior of the fibers with the aid of a theory (eg, Ref. 3 or 4). There is, however, a more direct technique for dealing experimentally with the determination of effective length, described by Drzal (eg, Ref. 20). The method consists of testing a composite comprising one fiber in a relatively large volume of matrix, in axial tension, and observing under a polarizing microscope the occurence and spacing of fiber breaks. This special uni-fiber composite is strained until no further fiber breaks occur; then the spacing between fiber breaks is readily interpreted to give estimates of effective length and interfacial shear strength. Currently, the method relies on transparency of the matrix for optical observations of fiber breaks, etc.. Application to opaque composites such as carbon-carbon would require additional development. At least two possible approaches might be attempted: destructive inspection of the specimens after test to measure the length of fiber fragments, or measurement (via Moire techniques or laser interferometry) and interpretation of surface strains to establish the spacing of fiber breaks during the test. Insofar as the Air Force Wright Aeronautical Laboratories currently is pursuing the application of the method to carbon-carbons (Ref. 21), we may anticipate interesting results.

The analysis presented here is relevant to multi-directional composites made with straight yarns. However, in a laminate or a 3D composite, failure of a bundle may not propagate to fail the composite, much as our tensile model shows that a single fiber break does not constitute failure of the bundle. Therefore, additional development of the tensile model may be necessary to deal with tensile fracture of multi-directional composites.

ILLUSTRATIVE RESULTS

For illustrative purposes, calculations have been done assuming σ_ℓ and τ' to be zero, and assuming the fiber strength distribution and the various thermal and elastic properties are independent of temperature. Table 2 shows the inputs and outputs of the analysis. Figure 4 is a plot of the results, in the same format as Figure 1. The results show that friction stress is dependent on yarn stress and temperature, and that the composite strength increases with temperature as is the case for real carbon-carbons (eg, Ref. 1).

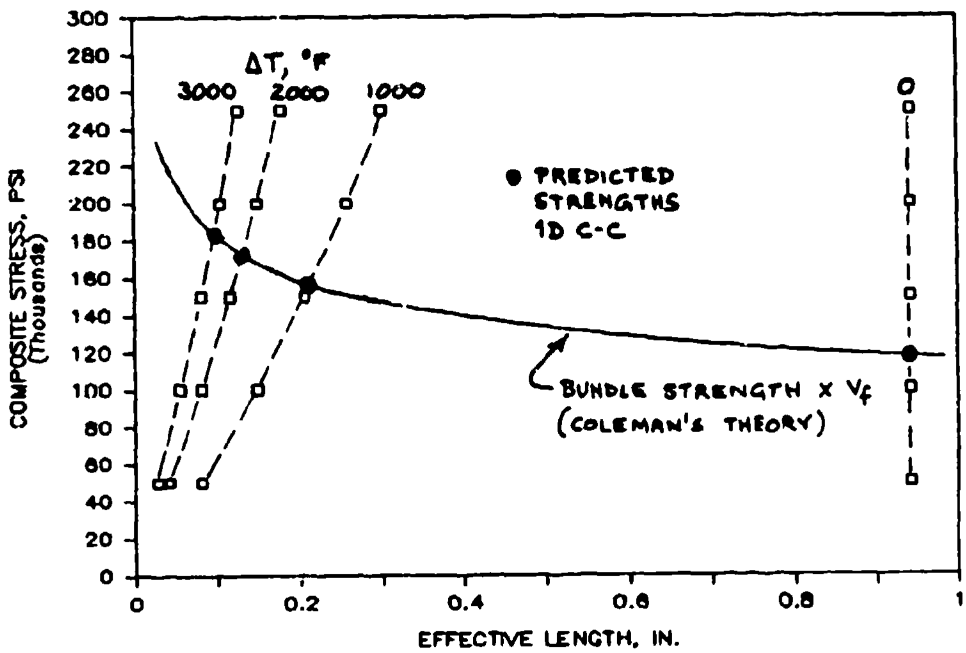


Figure 4. Illustrative results (see Table 2).

CONCLUSIONS

The model for tensile strength, presented here, estimates the expected strength of a uni-directional carbon-carbon as a function of the strength distribution of the fibers, the gage length of the composite test specimen, the residual thermal stresses at the fiber-matrix interface, the expected friction coefficient at the fiber-matrix interface, the transverse thermal expansion coefficients of the fiber and matrix, the stresses applied to the composite, and the Poisson's ratio of the yarn bundle.. Discrepancies between measured composite strength and the analytical predictions can, if these input properties are sufficiently well known, be attributed to fabrication-process induced changes in the strength distribution of the fibers.

By including transverse compression stress as a parameter, the model suggests that the tensile strength of yarn bundles in a multi-directional composite will be affected by mini-mechanical stresses and by transverse tractions applied to the composite.

By including the properties of the matrix (insofar as it affects the transverse properties of the uni-directional yarn bundle), and the frictional behavior of the fiber-matrix interface, the model may find use in guiding the selection of matrices and processes for improving tensile strength of carbon-carbon composites.

TABLE 2. ILLUSTRATIVE CALCULATIONS FOR UNIDIRECTIONAL C-C.

INPUTS (HYPOTHETICAL COMPOSITE, INDEPENDENT OF TEMPERATURE):

Weibull Modulus, m	5	for 1" long
Weibull Scale Fact., sig0	300000 PSI	fibers
Composite Axial Modulus, EcL	4.00E+07 PSI	
Avg Transverse Modulus, EbarT	4.00E+05 PSI	
Composite Poisson's Ratio, NUcLT	0.3	
Composite Expansivity, ALPcT	8.00E-06 per deg F	
Fiber Expansivity, ALPfT	1.00E-05 per deg F	
Fiber Volume Fraction, Vf	0.65	
Friction Coefficient, MU	0.3	
Fiber Radius, Rf	0.000138 inch	

OUTPUTS:

TEMP. RISE deg. F	ASSUMED COMPOSITE STRESS PSI	PREDICTED MAXIMUM FRICTION STRESS PSI	PREDICTED EFFECTIVE LENGTH IN.	THEORETICAL BUNDLE STRENGTH PSI
0	50000	11	0.9429	117081
0	100000	23	0.9429	117081
0	150000	34	0.9429	117081
0	200000	45	0.9429	117081
0	250000	56	0.9429	117081
1000	50000	131	0.0808	191372
1000	100000	143	0.1489	169361
1000	150000	154	0.2070	158561
1000	200000	165	0.2572	151825
1000	250000	176	0.3009	147126
2000	50000	251	0.0422	217911
2000	100000	263	0.0808	191372
2000	150000	274	0.1162	177953
2000	200000	285	0.1489	169361
2000	250000	296	0.1790	163228
3000	50000	371	0.0286	235608
3000	100000	383	0.0555	206337
3000	150000	394	0.0808	191372
3000	200000	405	0.1048	181692
3000	250000	416	0.1274	174717

The illustrative calculations have shown that increased strength at high temperatures may be expected for carbon-carbons, even if the fiber strength does not increase with temperature.

Given the expectation of weak fiber-matrix interfaces in carbon-carbons, the model suggests that carbon-carbon yarn bundles will be weaker than graphite-epoxy yarn bundles made with the same fiber, even if there is no degradation of fiber properties during making of the carbon-carbon composite.

Implementing the analysis requires much data that is currently unavailable or only poorly known. In particular, effort should be directed toward measuring transverse properties of fibers and interface strengths and frictional behavior. It is recommended that appropriate experiments be devised and conducted.

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REFERENCES

1. Chatterjee, S. N., McLaughlin, P. V., and Lowe, D. L., An Investigation of the Tensile and Shear Strengths of 3D Carbon-Carbon Composites, AIAA/ASME SDM Conference, 1976.

2. Manders, Peter W., Bader, Michael G., and Chou, Tsu-Wei, Monte Carlo Simulation of the Strength of Composite Fibre Bundles, Fibre Science and Technology, Vol.17, 1982, pp183-204.

3. Rosen, B. W., Tensile Failure of Fibrous Comopsites, AIAA J., Vol. 12, 1964, p1965ff.

4. Batdorf, S. B., A New Micromechanical Theory for Composite Failure, 1980 JANNAF RNTS Meeting, Monterey.

5. Phoenix, S. L., Probabilistic Strength Analysis of Fibre Bundle Structures, Fib. Sci. Tech., 1974, vol.7, p15ff.

6. Gent, A. N., Fracture Mechanics Applied to Elastomeric Composites, Rubber Chemistry and Technology, Vol.56, No.5, Nov-Dec 1983, pp1011-1018.

7. Den Hartog, J. P., Strength of Materials, Dover Press, 1949.

8. Reynolds, W. N., Physical Properties of Graphite, Elsevier, 1968.

9. Diefendorf, R. J., Proc. Conf. Carbon, 1959, p483ff and p489ff.

10. Jortner, J., Cracking in 3-D Carbon-Carbon Composites During Processing and Effects on Performance, Proc. Army Symp. Solid Mechanics, 1976, AMMRC MS 76-2, September 1976, pp81-97.

11. Evangelides, J. S., A Physically Based Analytical Tensile Model for Carbon-Carbon Composites, 13th Biennial Conf. Carbon, 1977, p351ff.

12. Evangelides, J. S., Microstructural Characterization of Crack Propagation and Analytical Modeling of Carbon-CArbon Composites, Proc. Army Symp. Solid Mech., AMMRC MS 76-2, Sept 1976.

13. Jortner, J., and Clayton, F. I., Carbon-Carbon Microstructure: Causes and Effects on Nosetip Ablation, McDonnell Douglas Report MDC G6390, April 1976.

14. Eitman, D. A., et al, Fiber Matrix Interactions in Carbon-Carbon Composites, 19th National Symposium SAMPE, April 1974.

15. Jortner, J., et al, In-Situ Densities of Fibers and Matrices in Carbon-Carbon Composites, McDonnell Douglas Report MDC-G7385, May 1978.

16. Albugues, F., et al, Influence of High Temperature Thermal Treatments on the Properties of Carbon Fibers, Extended Abstracts of 16th Bienniel Conference on Carbon, American Carbon Society, July 1983, pp492-493.

17. Seibold, R. W., et al, Advanced Densification of Carbon-Carbon Composites.... AFML-TR-79-4003, Feb. 1979.

18. Loomis, W. C., et al, Rocket Nozzle Processing Science (APIC/RN), AFWAL-TR-81-4118, November 1981.

19. Clayton, Fred I., Personal Communication, May 1984.

20. Drzal, Lawrence T., et al, Adhesion of Graphite Fibers to Epoxy Matrices, I. The Role of Fiber Surface Treatment, J. Adhesion, 1982, Vol. 16, pp1-30.

21. Drzal, L. T., Personal Communication, April 1984.

22. Sines, George, Personal Communications, 1984.

23. Jortner, J., Effects of Anomalies on Thermostructural Behavior of 3D Carbon-Carbon Composites, AFWAL-TR-82-4118, October 1982.